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SIMULATION OF CONTINUOUS CASTING

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SIMULATION OF CONTINUOUS CASTING

by

J. Bowker

ABSTRACT

A technique to simulate continuous casting involving melting, solidification and tensile testing at high temperatures (700-1100°C) using the Gleeble 1500 is described. Test results on the assessment of cracking susceptibility during continuous casting of four C-Mn steels are reported.

LABORATOIRES DE RECHERCHE NE METALLURGIE PHYSIQUE

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SIMULATION DE COULEE CONTINUE

par

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RESUME

Ce rapport décrit une technique de simulation de coulée continue incluant la fusion, la solidification et l'essai de tension à haute température (700 - 1100°), à l'aide du Gleeble 1500. Les résultats de l'évaluation de la susceptibilité à la fissuration lors de la coulée continue de quatre aciers C-Mn font partie intégrante du rapport.

INTRODUCTION

A greater understanding of the factors which contribute to cracking of continuously-cast slabs has been achieved by examining the hot ductility of steels subjected to a thermal cycle typical of that experienced in a given region of a slab during the casting process. The simulation of continuous casting has involved the melting, solidification and testing of tension specimens resistance heated in situ in a Gleeble 1500 weld simulator. This device[1] permits a wide range of controlled thermal cycling and mechanical loading. Effects of cooling rate from the melt, strain rate, test temperature and time at test temperature on hot ductility for a range of steel compositions have been examined by a number of workers[2-5] using this technique. In some instances the steel is given a "sensitizing" treatment immediately prior to testing instead of being melted. This consists of heating the tensile test sample to within 100°C of T_m, cooling it rapidly to the test temperature and testing immediately. This is considered equivalent to testing the steel as it cools following solidification. Suzuki et al.[2] identified three regions Of embrittlement using the two (above) simulation techniques. In region I between T_m and 1340°C poor ductility was found to be due to the presence of liquid films in the interdendritic regions. This observation was supported by Brimacombe and Sorimachi[3], who stated that liquid films apparently contained high levels of sulphur and phosphorus which segregated between growing dendrites. This embrittlement is strain rate insensitive and thought to be responsible for most of the internal cracks found in continuous casting.

A second region of embrittlement appears in the temperature range 1200 to 900°C and according to Suzuki[2] can be attributed to the precipitation of finely distributed oxy-sulphides at the austenite grain boundaries. When the steel has a sulphur content over 0.01% and a low Mn/S ratio there is marked embrittlement. Overaging treatments such as slow cooling, isothermal holding or a slow strain rate result in good ductility.

A third region of embrittlement is present between 900°C and 600°C and as in the case of embrittlement between 1200°C and 900°C has been shown to be associated with intergranular precipitation. The difference however is that the degree of embrittlement between 900°C and 600°C increases as the rate of deformation decreases.

The relative contributions of the two lower temperature regions of embrittlement towards cracking have been examined by Brimacombe and Sorimachi[3] and Weinberg[4]. Their results do not present a very clear picture of the failure mechanism, although Weinberg states that if bending of a continuous-cast slab occurs in a region of poor ductility then cracking can occur when sufficiently large tensile strains are imposed on the steel.

In order to examine the extent to which poor ductility exists at temperatures associated with the bending operation, a program was carried out to measure the high temperature ductility of a range of low-C steels following a melting simulation.

EXPERIMENTAL PROCEDURE

The compositions of four steels examined in this program are given in Table 1. A procedure of controlled melting, solidification and subsequent tensile testing described in a previous report[6] was used to examine hot ductility over the temperature range 700 to 1100°C. A strain rate of $5 \times 10^{-3} s^{-1}$ was employed during tensile testing which is within the range of bending strain rates present during continuous casting. Figure 1 represents the thermal/mechanical cycle to which samples were subjected. Tests were carried out for each condition to produce three results where the sample fractured at the location of the thermocouple. Following testing, samples were prepared for optical metallography by sectioning longitudinally, polishing and etching in 2% nital. Fracture surfaces were examined in a scanning electron microscope capable of energy dispersive X-ray analysis.

RESULTS AND DISCUSSION

Figure 2 shows the hot ductility curves for the 4 steels in the temperature range 700 to 1100°C. Steel B has the highest ductile to brittle transition temperature in the range 925 to 1000°C while for steel A the transition to brittle failure is at much lower temperature i.e., 850°C. Both steels C and D have much shallower and narrower ductility troughs with a minimum at 800°C in both cases. In the low ductility region all the steels failed intergranularly with ductile dimples present on the grain boundary facets (shown typically in Fig. 3). The scanning electron microscope identified the

presence of small particles, rich in Mn and S, less than 1 μm in diameter in most of the dimples. The following mechanism has been proposed by Suzuki et al.[2] to explain embrittlement at these lower temperatures. When a tensile stress is imposed on carbon steels with sulphides, carbides or nitrides precipitated at austenite grain boundaries, these precipitates become sources of stress concentration, causing the formation, growth and coalescence of voids. Formation of ferrite along the grain boundaries and the difficulty of recovery and recrystallization of the austenite phase at lower temperatures acts to localize deformation at the grain boundaries. This point is supported by the fact that the tensile strength of the grain boundary ferrite is several times smaller than that of austenite at the same temperature.

In the present study further work is necessary to characterize the precipitates at the prior austenite grain boundaries and to determine whether they are present above the ductile-brittle transition temperature. This would indicate whether the drop in ductility is associated with their formation or whether there are other contributory factors. In the mechanism reported by Suzuki it is suggested that the onset of brittle behaviour is linked to the formation of grain boundary ferrite. A series of quench experiments from various testing temperatures failed to produce results consistent with this argument. For example steel B tested at 900°C showed marked embrittlement whereas the microstructure of a sample quenched from that temperature failed to reveal any significant grain boundary ferrite. On the other hand steel A which is ductile at 900°C contained a high proportion of grain boundary ferrite following quenching (see Fig. 4). Thus although intergranular precipitation of sulphides, of which the formation mechanism is not fully understood, does contribute to embrittlement, further study is needed to explain the differences in hot ductility observed in this work.

In a qualitative sense the results of this study show that if the continuous-casting bending operation were to be carried out at 950°C for example, steel B, which has the poorest ductility at this temperature, would be the most susceptible to cracking.

CONCLUSIONS

1. A technique to simulate continuous casting involving melting, solidification and tensile testing at high temp (700-1100°C) using the Gleebe 1500 was successfully employed in an assessment of the cracking susceptibility during continuous casting of four C-Mn steels.
2. All four steels exhibited embrittlement in the temperature range 700 to 1100°C characterized by an intergranular failure with ductile dimples on the grain boundary facets. Mn and S rich precipitates were associated with the locally ductile behaviour.
3. Steel B showed the highest ductile to brittle transition temperature at approximately 1000°C indicating a lower susceptibility to cracking during the bending operation than the remaining three steels whose ductile-brittle transition temperatures were less than 900°C.

ACKNOWLEDGEMENTS

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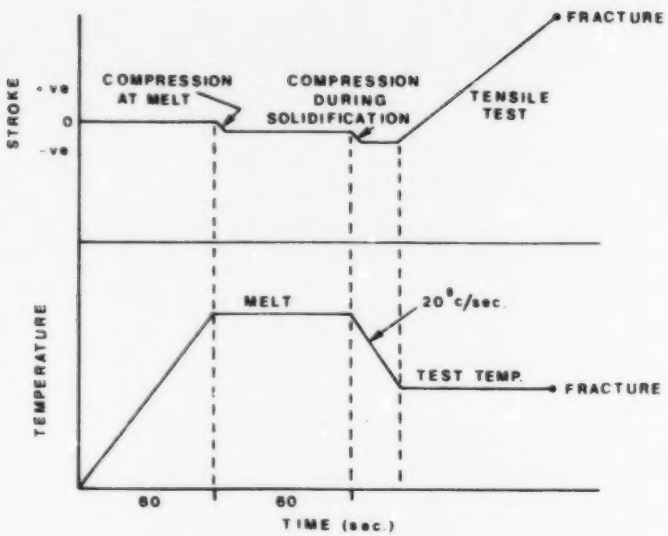


Fig. 1. Thermal and mechanical program for melting and hot ductility testing. (neg 13525)

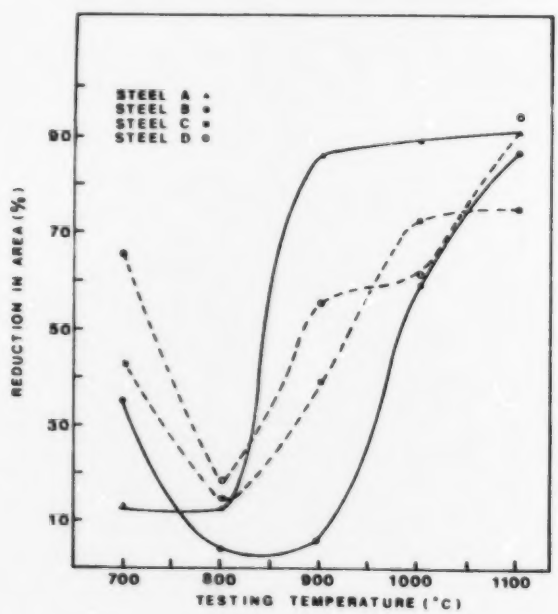
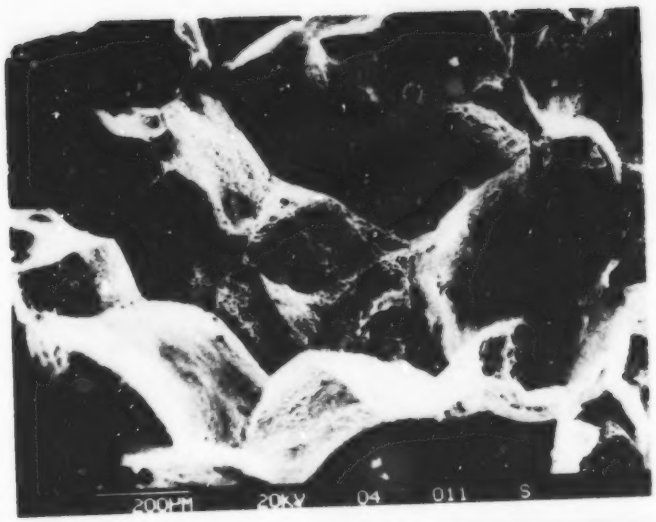
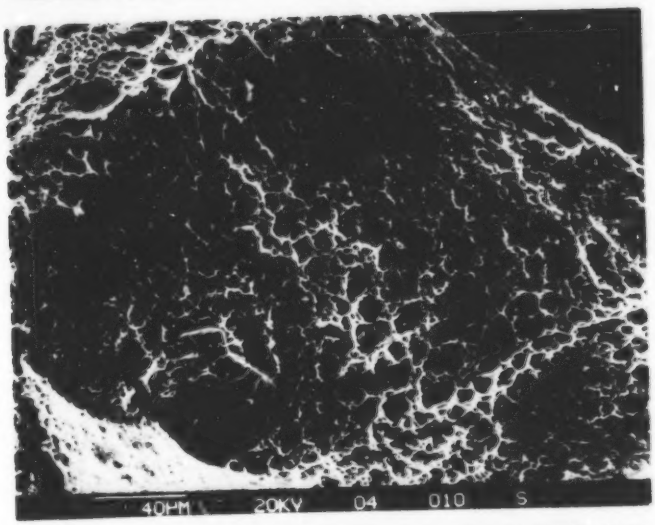


Fig. 2. Hot ductility curves. (neg 13524)

(a) (neg 13654)



(b) (neg 13652)



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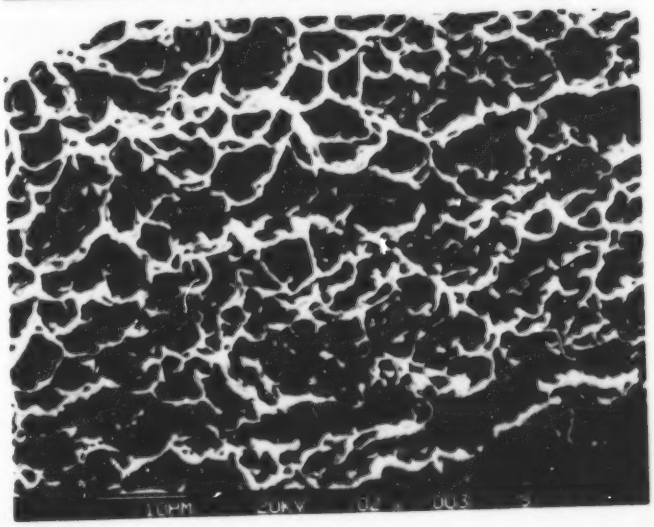

Fig. 3. Scanning electron micrographs of fracture surfaces of Steel C following testing at 800°C.

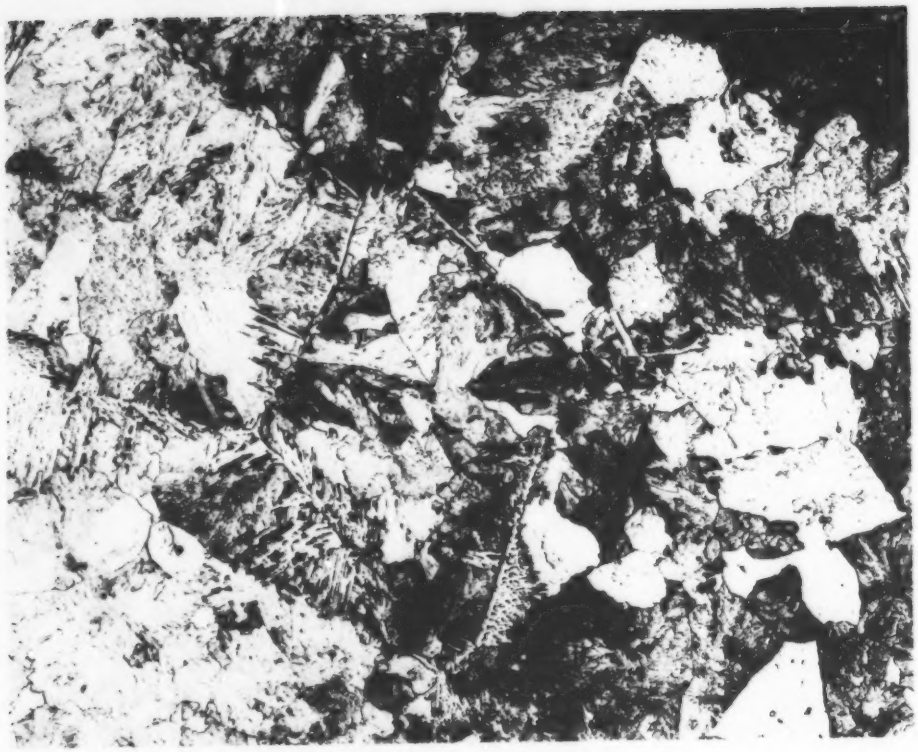

Fig. 4. Optical micrograph of Steel A quenched from 900°C following melt simulation. (neg 13651)